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ALUMINUM ALLOY CASTINGS - PRODUCTION ROUTES FOR OPTIMUM FRACTURE TOUGHNESS

Report No 87-172-K

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ABSTRACT

The importance of pouring temperature and cooling rate through the freezing range have been established and quantified for optimization of tensile and fracture toughness properties in a number of commercially popular cast aluminum alloys. Production techniques developed are applicable to processes using sand, plaster or permanent metal moulds. Soundness and fracture toughness have been greatly improved with no sacrifice in tensile properties.

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EXECUTIVE SUMMARY

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Most specifications for aluminum alloy castings are particularly undemanding. Despite this many suppliers are unable to meet minimum ductility requirements and the fracture toughness of their products is extremely poor. A research program was carried out to determine why the castings perform so poorly and to demonstrate how considerable improvements may be obtained. This report defines the impact and importance of alloy composition, melting technique, pouring temperature, mould material and temperature and heat treatment in provision of high integrity castings for transmission related structures. The successful techniques developed are applicable to processes involving use of sand, plaster or permanent metal moulds.

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ALUMINUM ALLOY CASTINGS - PRODUCTION ROUTES FOR OPTIMUM FRACTURE TOUGHNESS

1.0 INTRODUCTION

Many cast aluminum alloy components are utilized in the construction, operation and maintenance of transmission and distribution structures. Typical examples are overhead conductor suspension clamps and grounding cable clamping and connecting devices. Unfortunately specifications covering cast aluminum components are typically undemanding on requirements for tensile ductility and the question of fracture toughness is not addressed. Also in most cases specified tensile properties need only be met by separately cast non-machined test bars which rarely reflect the properties of actual production castings. Consequently, the customer receives castings which often experience brittle fracture under the influence of normal installation stresses.

It is encouraging to note that ASTM now recognize this problem and have introduced a more demanding specification, ie B686 which strongly recommends the use of chills in sand castings and demands somewhat higher tensile ductility values in specimens machined from actual sand castings. Our research work intended to progress further in this desirable direction and to determine control parameters for production of optimum tensile ductility and fracture toughness in cast aluminum alloy components.

2.0 HISTORICAL BACKGROUND TO THE PROBLEM

Because of their excellent fluidity and good overall foundry handling qualities the alloys most favoured by the suppliers are based on the aluminum-silicon binary system. The heat-treatable 7 wt% silicon alloy 356 or its premium quality variant A356 represents the most favoured composition. The requirements of various specifications for the premium grade are listed in Tables 1 and 2.

Where considerations of possible warpage apply or a requirement for optimal tensile ductility exists, the manufacturer will typically utilize a non-heat treatable alloy from the aluminum-magnesium system. Tables 1 and 3 present specified requirements for the commonly chosen type 535.0 alloy.

Regardless of the type of alloy selected, it has been repeatedly demonstrated that tensile properties of specimens machined from failed castings fall considerably below the very modest minimum values tabulated in the purchase specifications.

Both of the aforementioned popular alloys have wide freezing ranges, ie 56°C for A356 and 80°C for type 535.0. Consequently, when foundry practice results in relatively slow cooling through the casting solidification range there is a marked tendency toward development of scattered interdendritic porosity in the aluminum-silicon eutectic type system and gross intergranular unsoundness in the peritectic type aluminum-magnesium alloy system. The stress concentrating effects and loss of load bearing area resulting from such discontinuities contribute substantially toward the problem of brittle fracture under very modest loading conditions. Also resulting from slow solidification rates is the excessive growth of coarse microstructural constituents. When the brittle silicon phase in alloy A356 exists as a semi-continuous network of large platelets the overall toughness of the casting is greatly compromised. Whilst the matrix properties of the 535.0 alloy are generally acceptable in terms of tensile strength, ductility and toughness the overall determinant of casting behaviour is usually the degree of grain boundary unsoundness which is generally sufficiently severe to guarantee brittle fracture.

2.1 The Influence of Foundry Practice on Casting Quality

Attainment of a suitably high solidification rate has been stressed as crucial if widely scattered defects attributable to either interdendritic or intergranular shrinkage are to be avoided. Attendant benefits to the overall microstructure of the aluminum-silicon alloys were also discussed. Further refinement of the silicon phase is possible by incorporation into the alloy composition of a trace (0.005 - 0.015 wt%) amount of a suitable modifying element. Sodium has been the most widely used with strontium as a popular alternative and antimony showing some interesting possibilities.

Rapid solidification clearly depends on heat being swiftly conducted from the mould. With permanent metal moulds their high conductivity favours this desirable objective. Sand and to an even greater degree plaster have highly insulating properties and it is generally necessary to incorporate strategically located metal chills to act as heat sinks. Solidification rates in the vicinity of chills are high and undercooling below the alloy solidus avoids scattered shrinkage problems. The judicial placement of chills is designed to provide both a sound structure in critical areas of the casting and to favour progressive planar (ie non-dendritic) solidification towards

risers and feeders, thereby confining shrinkage to disposable appendages beyond the boundaries of the casting proper.

Undercooling and rapid planar solidification in a controlled direction also depend upon establishment and maintenance of steep thermal gradients. Toward this end alloy pouring temperatures and mould surface temperatures should be as low as possible and in the case of the mould should ideally remain so throughout the solidification process.

Most of our suppliers use either oil or gas fired furnaces of the crucible or reverberatory type. Significant disadvantages are inherent in the use of such equipment. Slow melting rates and in the case of reverberatory furnaces, particularly long holding times both result in unacceptable loss of modifying elements by volatoization. This is most serious when sodium is employed. Long term exposure to the atmosphere and combustion products introduces undesirable contaminants which may result in dross entrainment or gas porosity in finished castings. Degassing often removes modifying elements necessitating further ladle modification additions. Finally, difficulty in temperature control results in variable pouring conditions which adversely affect production of sound castings. Use of medium frequency electric induction furnaces avoids all of the aforementioned difficulties and provides not only uncontaminated compositionally homogeneous alloy of precise pouring temperature but also results in a clean, cool, quiet working environment and high productivity.

The desire to minimize thermal shock and attendant possible premature failure of permanent metal moulds may prompt a suppplier to use an excessively high pre-heating temperature which has an adverse effect on establishment and maintenance of desirable temperature gradients during casting solidification. Non-metallic moulds and generously proportioned strategically located associated metal chills should be held at a temperature no higher than that necessary to maintain safe pouring conditions by expulsion of moisture.

3.0 EARLY RESPONSES TO THE LOW STRENGTH & EMBRITTLEMENT PROBLEM

Recognizing the variations in technical sophistication among our suppliers, the prohibitive cost of permanent metal moulds for small orders and a general reluctance to use chills in non-metallic moulds, a decision was made to specify only alloys with very small freezing ranges. These are not susceptible to development of widely scattered shrinkage defects. Unfortunately, available alloys of this type fall into two distinct groups each of which has its own disadvantages. The simple non-heat treatable binary eutectic type such as the aluminum 10-12 wt% silicon composition is purchased to specifications guaranteeing tensile yield strength which may be as low as

60 MPa. Higher strength precipitation hardened aluminum-silicon alloys of approximately eutectic composition are available but their specified minimum tensile elongation values are less than 1% on a gauge length of 4 A and they tend to be extremely brittle. Despite the foregoing, the low strength eutectic alloy has proved to be very useful in certain applications.

3.1 Rationale for Possible Applicaton of a Low Strength Eutectic Alloy

Figure 1 illustrates failure by brittle fracture of a typical switchgear component. The component was originally fabricated in alloy 535.0 and supplied to ASTM specification B26 for sand castings. Considering the loss of load bearing area represented by the grain boundary shrinkage defects of Figure 2, it is not surprising that the casting failed prematurely under normal loading conditions. A replacement sand cast in alloy 356-T6 fared no better. Hardness tests on this material gave results which correlated well with acceptable tensile strength values. However, as noted in Figures 3 to 6, the loss of load bearing area due to the presence of interdendritic shrinkage porosity is considerable. Measurement of this area on the fracture face of the broken tensile test specimen of Figure 6 allowed a correction factor to be applied to tensile strength results which then placed the ultimate tensile strength in excellent correlation with the hardness of the alloy.

Attempts to eliminate the scattered shrinkage pores by having recourse to an alloy of near eutectic composition were successful. A desire to maintain yield and ultimate tensile strength properties at or above the minima specified for the previously used 535.0 and 356-T6 alloys resulted in selection of alloy 336.0 - T65. This alloy is normally poured into a permanent metal mould and there is no specified minimum tensile elongation value. Sand castings in this heat treated alloy had the desired tensile strength but essentially zero ductility and very poor fracture toughness. Brittle fracture resulted from normal bolt tightening stress during component installation.

Considering the evidence displayed in Figure 6 it is clear that a casting which loses half of its load bearing area due to interdendritic shrinkage porosity is no better than one having half the material yield strength and no porosity. Using this rationale a number of successful substitutions have been made using the relatively tough and ductile aluminum-silicon non-heat treated eutectic composition alloy S12N ordered to CSA specification HA9. A related significant advantage of this alloy is excellent weldability with no tendency toward development of heat-affected zone unsoundness.

4.0 DEVELOPMENT OF A PROGRAM FOR OPTIMIZATION OF DUCTILITY AND FRACTURE TOUGHNESS IN HIGH STRENGTH ALUMINUM ALLOY CASTINGS

Preliminary responses to unacceptably high failure rates had noted the suppliers tendencies to use high freezing range alloys and to utilize foundry techniques which invariably resulted in gross and widespread interdendritic porosity or intergranular unsoundness. Compounding the problems of stress concentration and loss of load bearing area introduced by these defects was a general failure to provide a suitably modified tough, ductile matrix microstructure. Substitution with low freezing range alloys of near eutectic composition was successful in eliminating unsoundness and where low yield strength was acceptable this approach provided an effective solution to the problem of brittle fracture. However, for many applications a combination of both high tensile strength and superior fracture toughness is preferred. A research program was devised to determine the production parameters necessary to achieve this goal. The controllable variables in the casting process are alloy composition; melting technique; pouring temperature, mould material and temperature and heat treatment parameters.

A decision was made to utilize alloy compositions most favoured by our suppliers. These are listed in Table 1 as alloys A356 and 535.0. The benefit of also including a low freezing range alloy of eutectic compostion was recognized and a strontium modified version of the West German DIN 1725 specification for alloy wa/3.2381.61 was selected. Compositional details of this sand casting alloy and a related one for permanent metal mould casting are included in Table 1.

Available equipment dictated that melting would be carried out in electric resistance heated refractory crucibles. A 0.5 kg charge could be heated to the desired pouring temperature in 30 minutes. The lowest feasible pouring temperature was considered to be 50°C above the liquidus. Pouring temperatures of 700°C and 800°C were used to assess the effect of this variable on casting properties. The moulds had tapered cylindrical cavities 132 mm high. Diameters were 31.5 mm at the top and 25.4 mm at the bottom. Mould materials were selected to offer a wide range of thermal conductivities and heat transfer properties. In descending order of efficacy the following were selected:

a) Copper;
b) AISI type 1020 carbon steel;
c) Plaster of Paris with a 52 mm thick copper chill in the base; and
d) Plaster of Paris without the chill.

Sand moulds for an actual production part were obtained from a casting supplier. Details of the part appear in Figure 7.

Laboratory produced heat treatable castings were water quenched and artificially aged to the T6 condition.

4.1 Effect of Mould Material

Cavities of the aforementioned shape and size were machined in copper and steel barstock of 75 mm diameter and 150 mm height. Plaster moulds with and without chills were of the same shape and size.

Chromel-Alumel thermocouples were positioned along the centre axes of the mould cavities with their hot junctions located 25 mm from the bottom. For the commercially fabricated sand mould of the air-set type a similarly dimensioned feeder cavity was selected to house the thermocouple. The walls of this cavity were backed with 150 mm to 450 mm of sand. To expel moisture and thereby ensure safe pouring conditions the moulds were heated to 100° prior to pouring. Figure 8 provides cooling curves and alloy solidification rates obtained with the various mould materials. The curves were obtained by monitoring the cooling of type A356 alloy which was held at 700°C for 10 minutes prior to pouring. Similar solidification rates were obtained with the type 535.0 and DIN 1725 alloys. The heat treatable alloys were solution treated and aged to the T6 condition. Hounsfield No 14 tension test coupons and standard 10 mm x 10 mm un-notched Charpy test bars were machined with their longitudinal centre axes approximately coincident with casting section mid-radial positions. Tables 2 to 4 compare test results with those obtained from material removed from commercial castings. As anticipated there is a general trend showing marked improvement in mechanical properties with increasing solidification rate. A critical value was achieved with steel moulds, therefore the even faster cooling rate achieved with copper produced essentially no further improvement.

4.2 Effect of Pouring Temperature

Clearly this variable has the potential to influence solidification rates and associated mechanical properties in the aluminum casting alloys under investigation. Data presented in Table 5 show this effect in alloy A356 poured into metal moulds. Reduction in solidification rate occasioned by an increase in pouring temperature from 700 to 800°C was minor. The effect on mechanical properties was negligible. However, it may be confidently anticipated that much higher pouring temperatures would have a more deleterious effect on alloy strength and toughness.

4.3 Effect of Mould Temperature

For rapid alloy solidification it is essential that a steep temperature gradient be maintained between the mould material and the cast alloy. This is difficult to achieve in sand and plaster moulds without recourse to metal chills. Extending the life of permanent metal moulds by use of excessive

preheat to minimize thermal shock cycle damage may be expected to adversely affect casting solidification rates. The effect is significant as evidenced by results appearing in Tables 5 and 6.

4.4 Effect of Alloy Composition

Data displayed in Table 4 indicates that early efforts to produce tough and ductile high yield strength castings in low, freezing temperature range heat treatable aluminum-silicon-magnesium alloys were only partially successful. At a yield strength of 235 MPa, tensile elongation values of 4% and unnotched Charpy impact results of 9J absorbed energy represented only a modest combined strength and toughness improvement over properties obtainable from the non-heat treated eutectic composition aluminum-silicon alloy S12N sand cast to CSA specification HA9. Since the increased strength and accompanying reduction in toughness and ductility are attributed to the precipitation hardening potential of magnesium combined with silicon in the heat treatable alloy, it was considered useful to reduce the magnesium concentration by approximately 50%. Results presented in Table 5 indicate that with a magnesium concentration of 0.22 wt % tensile strength values are maintained whilst ductility and toughness properties are substantially improved.

4.5 Correlation between Mechanical Properties and Microstructure

A strong correlation has emerged between alloy solidification rate and resultant strength and toughness properties. Microstructural features responsible for these developments are illustrated in Figures 9 and 10. Solidification times of less than 0.25 minutes in permanent metal moulds produced relatively large volumes of aluminum-silicon eutectic in which the hard silicon phase was present as small well dispersed spheroids or short rounded rod-like particles. Details are displayed in Figures 9a and 10a respectively for the 7 wt % silicon A356 alloy and the near eutectic composition wa/3.2381. Plaster moulds with a copper chill produced less pronounced undercooling in the poured alloys. Consequently, the microstructure in Figure 9b has a lower, nearer equilibrium concentration of the eutectic phase. The microstructural effects of undercooling were less pronounced with the strontium modified near eutectic alloy as illustrated by a comparison of Figures 10a and b. Eutectic concentration together with silicon particle shape, size and distribution are similar in castings from both permanent metal moulds and plaster moulds with metal chills. Casting into sand or plaster without chills led to the development of very coarse eutectic microstructures as observed in Figures 9c and d together with 10c and d. From Figures 9 and 10 together with Tables 2 and 4 a clear correlation emerges in which for a given tensile strength level toughness and ductility increase markedly with eutectic refinement.

A comparison between Figures 2a and 11 shows the effect of increased solidification rate on the micostructure of alloy 535.0. The commercially sand cast alloy has a relatively coarser primary solid solution grain size and shows a marked tendency toward development of grain boundary shrinkage cavities in poorly fed $Al/Mg_2 Al_3$ peritectic solidification zones. A dark etching microconstituent $Fe Al_3$ is also visible in Figures 2a and 11. The latter Figure demonstrates the effectiveness of undercooling and controlled thermal gradient development in both refining the primary phase grain size and eliminating grain boundary unsoundness. The benefits show in Table 3 as a markedly improved toughness level with no sac'ifice in either tensile strength or elongation values.

5.0 CONCLUSIONS

1. Commercial suppliers of aluminum alloy castings for application in electrical transmission line components frequently encounter difficulties in meeting the very modest mechanical property requirements of specifications such as ASTM B26 and B108 for sand and permanent mould castings respectively. Associated with marginal tensile strength and low ductility the castings also display very poor fracture toughness properties.

2. Poor performance results from stress concentrating interdendritic or in some cases intergranular shrinkage defects and the coarse lenticular form of brittle secondary phases. These can be avoided by control of alloy composition; pouring tmeperature, mould material, mould temperature and heat treatment. For sections up to approximately 30 mm thick this work has identified the critical parameters and demonstrated their effectiveness in producing sound, strong and tough components in alloys A356, 535.0 and wa/3.2386.

3. To achieve the desired properties, pouring temperature may be between 700 and 800°C. The temperature of either refractory or permanent metal moulds should be controlled in the range of 100-125°C. Use of metal chills (eg copper) is essential when using sand or plaster moulds.

4. Although the introduction during 1985 of ASTM specification B686 is a welcome step in the right direction, it is still relatively undemanding and does not address fracture toughness. Based on the current work it would seem not unreasonable to draft specifications requiring minimum properties outlined in Table 7. These should be obtained on test specimens machined from actual castings.

ACKNOWLEDGEMENTS

Casting, heat treatment, mechanical testing and metallographic work was carried out by Mr. T.R. Ryans. We are also indebted to a number of suppliers for useful discussions and provision of commercial scale sand moulds.

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TABLE 1 - SPECIFIED COMPOSITIONAL REQUIREMENTS

Specification	Alloy	Element - wt % Concentration						
		Si	Fe	Cu	Mn	Mg	Zn	Ti
ASTM B26-86 &	A356	6.5-	0.20	0.20	0.10	0.25-	0.10	0.20
ASTM B108-85a	A356	7.5	max	max	max	0.45	max	max
DIN 1725	wa/	9.0-	0.50	0.05	0.40	0.20-	0.10	0.15
G Al Si 10 Mg	3.2381.61	11.0	max	max	max	0.50	max	max
DIN 1725	wa/	9.0-	0.60	0.30	0.40	0.20-	0.10	0.15
GK Al Si 10 Mg	3.2381.62	11.0	max	max	max	0.50	max	max
Hydro	Low Mag	10.2-	0.60	0.10	0.10-	0.23-	0.10	0.05
		12.2	max	max	0.30	0.27	max	max
ASTM B26-86	535.0	0.15	0.15	0.05	0.10-	6.2-	0.05	0.10-
		max	max	max	0.25	7.5	max	0.25

TABLE 2 - MECHANICAL PROPERTIES OF A356-T6 ALLOY CASTINGS

Mould	Source	UTS MPa	0.2% PS MPa	% El on $4\sqrt{A}$	% RA	Hardness HV	Un-notched Charpy	
							J	% LE
Steel	Commercial	240	175	4	-	80	6	2.0
Steel	Laboratory	311	200	20	20	77	85	20.3
Plaster & Chill	Laboratory	266	179	25	18	66	37	10.7
Sand	Commercial	171	171	2	-	99	2	0.1
Sand	Laboratory	240	166	6	5	60	9	3.0
Plaster	Laboratory	174	168	2	3	45	3	0.1
ASTM Spec	B26-86	234 min	166 min	3.5 min	-	-	-	-
ASTM Spec	B108-85a	228 min	179 min	5.0 min	-	-	-	-
ASTM Spec	B686-86	262 min	193 min	5.0 min	-	-	-	-

TABLE 3 - MECHANICAL PROPERTIES OF TYPE 535.0 ALLOY CASTINGS

Mould	Source	UTS MPa	0.2% PS MPa	% El on $4\sqrt{A}$	% RA	Hardness HV	Un-notched Charpy J	Un-notched Charpy % LE
Sand	Commercial	224	125	13	10	80	29	9.7
Steel	Laboratory	289	140	14	10	72	98	20.3
ASTM Spec	B26-86	241 min	124 min	9 min	-	-	-	-
ASTM Spec	B108-85a	241 min	124 min	8 min	-	-		-

TABLE 4 - MECHANICAL PROPERTIES OF ALLOY wa/3.2383.61/62-T6 ALLOY CASTINGS

Mould	Source	UTS MPa	0.2% PS MPa	% El on $4\sqrt{A}$	% RA	Hardness HV	Un-notched Charpy J	Un-notched Charpy % LE
Steel	Commercial	240	210	1	-	90	5	<1.0
Steel	Laboratory	278	235	4	3	69	9	2.3
Plaster & Chill	Laboratory	228	178	4	2	56	6	1.0
Sand	Commercial	260	200	4	3	102	3	<1.0
Steel*	Laboratory	285	196	9	6	70	25	5.1
DIN Spec 1725 GAl Si 10 Mg (sand castings)		220 min	180 min	1 min	--	80-100	-	-
DIN Spec 1725 GK Al Si 10 Mg (permanent metal mould)		240 min	210 min	1 min	--	85-115	-	-

* Magnesium concentration reduced from 0.50 wt % to 0.22 wt % in the wa/3.2383.61-T6 alloy.

TABLE 5 - ALLOY SOLIDIFICATION KINETICS

Alloy	Freezing Range (°C)	Mould Material	Pouring Temp (°C)	Mould Temp (°C)	Cooling Rate Through the Solidification Range °C per min	Solidification Time - mins
A356	56	Copper	700	100	467	0.12
A356	56	Copper	800	100	431	0.13
A356	56	Copper	800	350	233	0.24
A356	56	Steel	700	100	258	0.22
A356	56	Steel	800	100	243	0.23
S356	56	Steel	800	350	78	0.72
A356	56	Plaster & Chill	700	100	47	1.19
A356	56	Sand	700	100	25	2.24
A356	56	Plaster	700	100	14	4.00
wa 3.2383	8	Steel	700	100	-	0.17
wa 3.2383	8	Plaster & Chill	700	100	-	1.70

TABLE 6

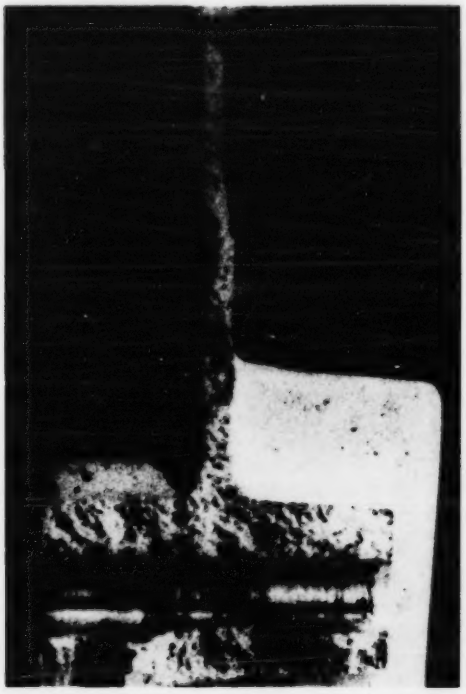
EFFECT OF PERMANENT METAL MOULD TEMPERATURE ON MECHANICAL PROPERTIES OF ALLOY A356-T6

Pouring Temp(°C)	Mould Material	Mould Temp(°C)	UTS MPa	0.2% PS MPa	% El on $4\sqrt{A}$	Un-notched Charpy J	Un-notched Charpy % LE
800	Copper	100	302	215	16	59	15.7
800	Copper	350	222	186	5	30	7.6
800	Steel	100	311	200	20	85	20.3
800	Steel	350	245	173	7	24	7.1

TABLE 7

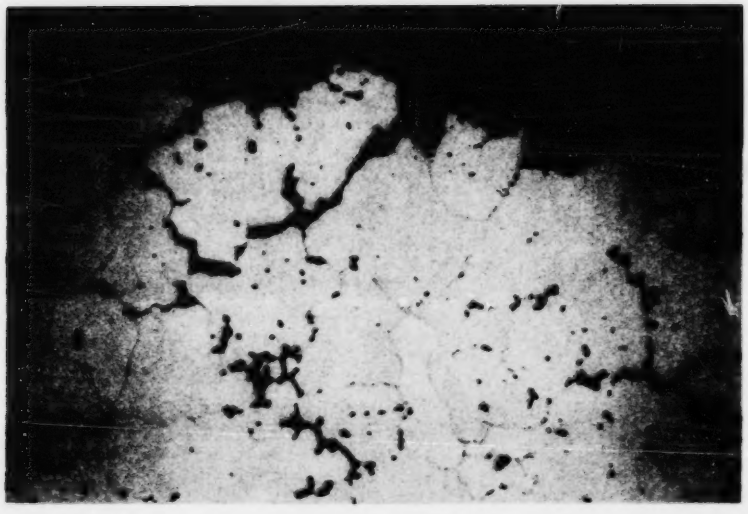
SUGGESTED SPECIFICATIONS - MECHANICAL PROPERTIES

Alloy	UTS MPa	0.2% PS MPa	% El on $4\sqrt{A}$	Hardness HV	Un-notched Charpy J	LE
A356-T6	250 min	170 min	10 min	61-77	25 min	7 min
535.0	241 min	124 min	10 min	59-74	25 min	7 min
Hydro Low Mag -T6	225 min	170 min	7 min	55-75	20 min	5 min



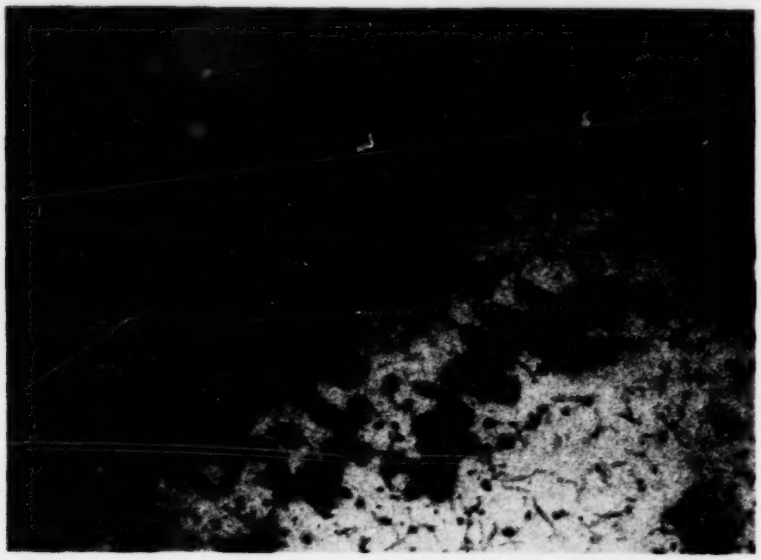
X 0.25 mag

FIGURE 1 - BRITTLE FRACTURE OF AN ALLOY 535.0 SAND CAST SWITCHGEAR COMPONENT



X 50 mag

FIGURE 2 - SAND CAST ALLOY 535.0 WITH GROSS GRAIN BOUNDARY UNSOUNDNESS

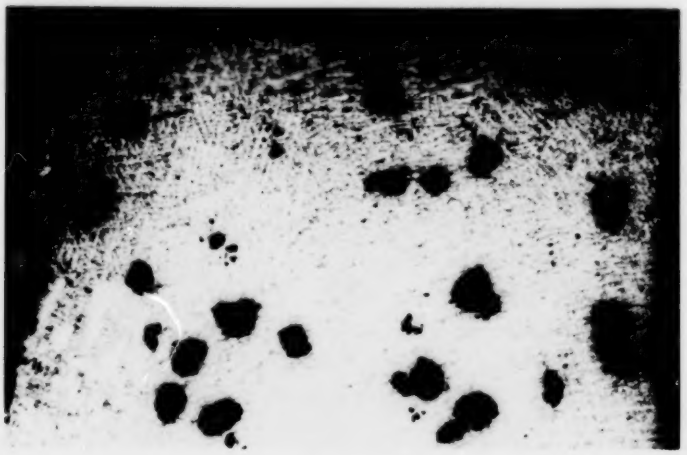


X 50 mag

FIGURE 3 - GROSS INTERDENDRITIC SHRINKAGE POROSITY IN SAND CAST ALLOY 356-T6

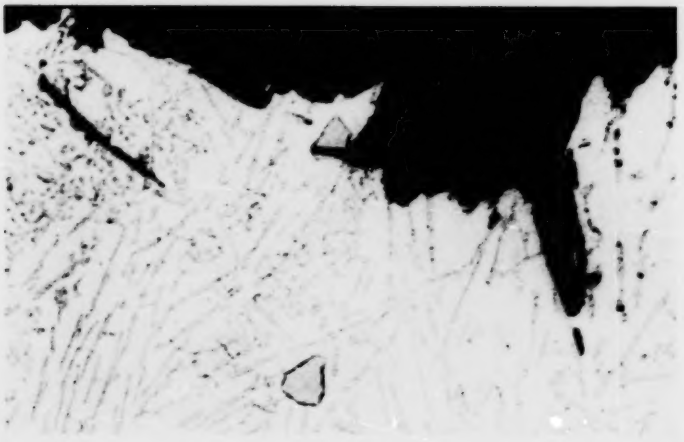
X 100 mag

FIGURE 4 - GROSS INTERDENDRITIC SHRINKAGE POROSITY IN SAND CAST ALLOY 356-T6



X 3.5 mag

a) Interdendritic Shrinkage Porosity



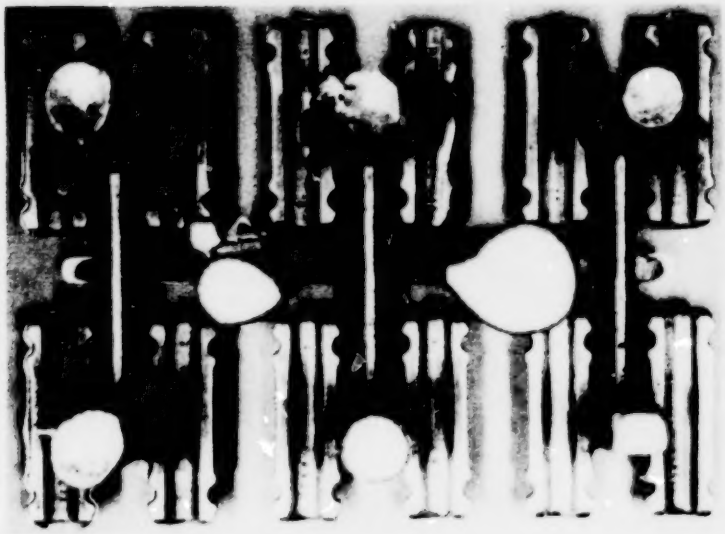
X 200 mag

b) Coarse Eutectic Structure

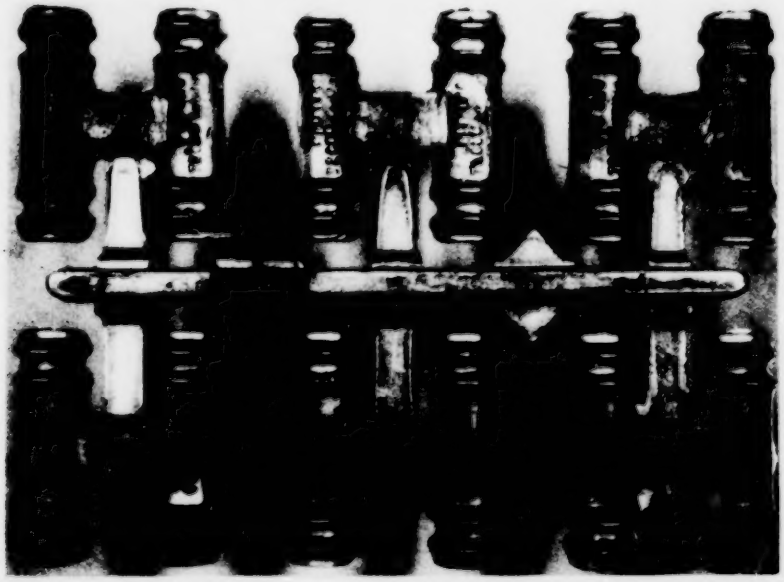
FIGURE 5 - INTERDENDRITIC POROSITY AND COARSE EUTECTIC STRUCTURE IN SAND CAST ALLOY 356-T6

X 15 mag

FIGURE 6 - TENSILE TEST SPECIMEN SHOWING SUBSTANTIAL LOSS OF LOAD BEARING AREA DUE TO INTERDENDRITIC SHRINKAGE POROSITY

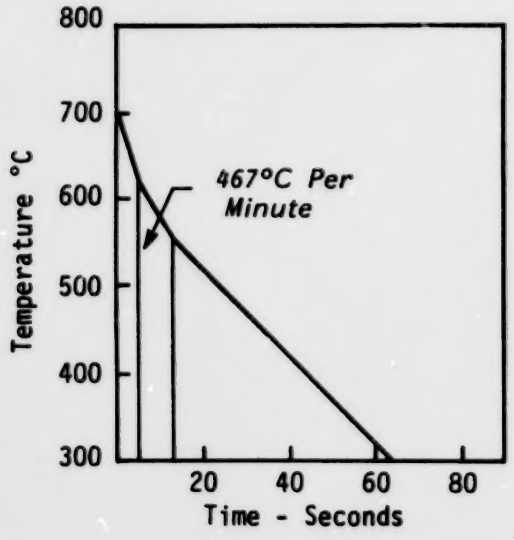


a) Plan View X 0.15 mag

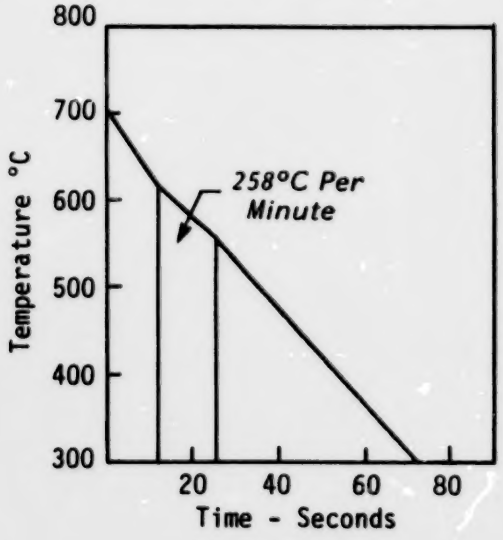


b) Underside X 0.17 mag

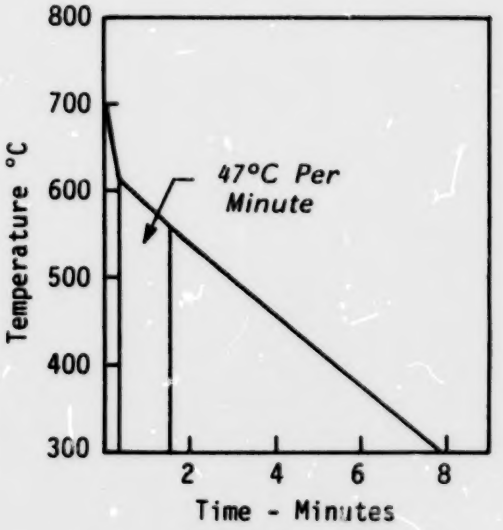
FIGURE 7 - DETAILS OF THE GATING, RISERING & CASTING LAYOUT FOR A CABLE CLAMP COMPONENT



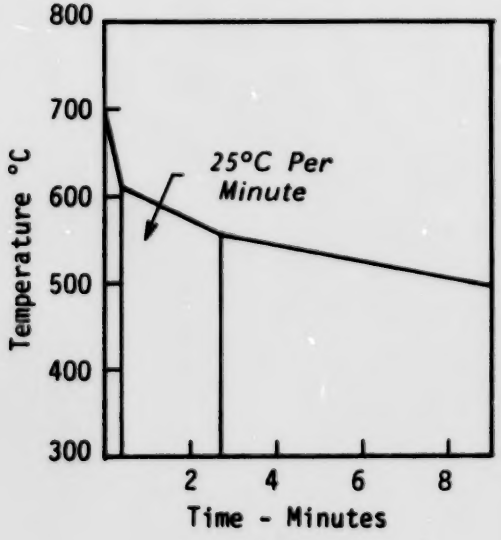
a) Copper



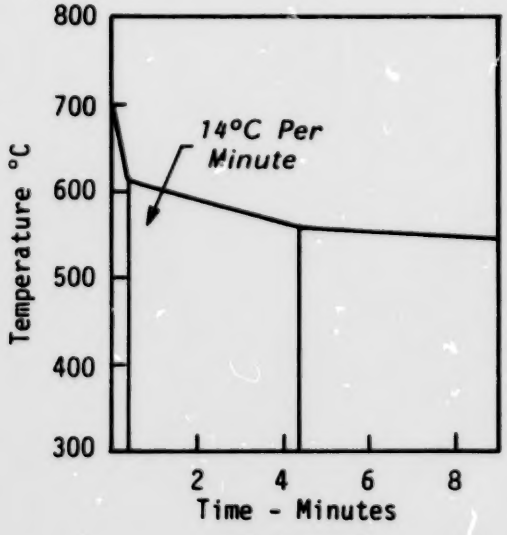
b) Steel



c) Plaster + Chill



d) Sand

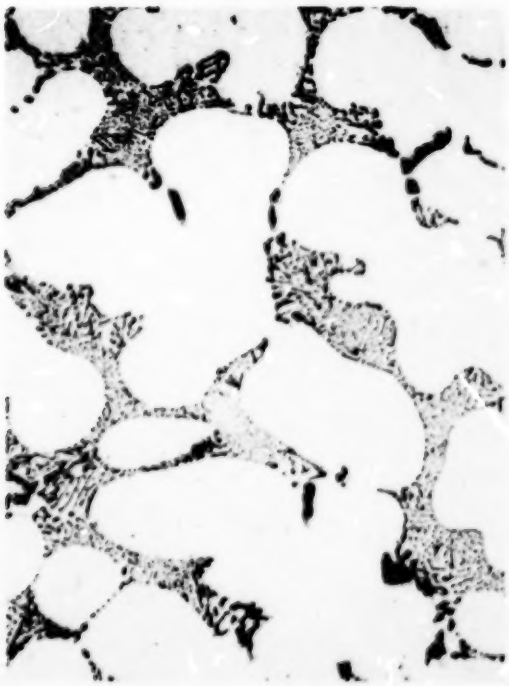


e) Plaster

FIGURE 8
COOLING CURVES OBTAINED DURING SOLIDIFICATION OF ALLOY A356

175266-RD
87-172

a) 258°C per min
(Steel Mould)

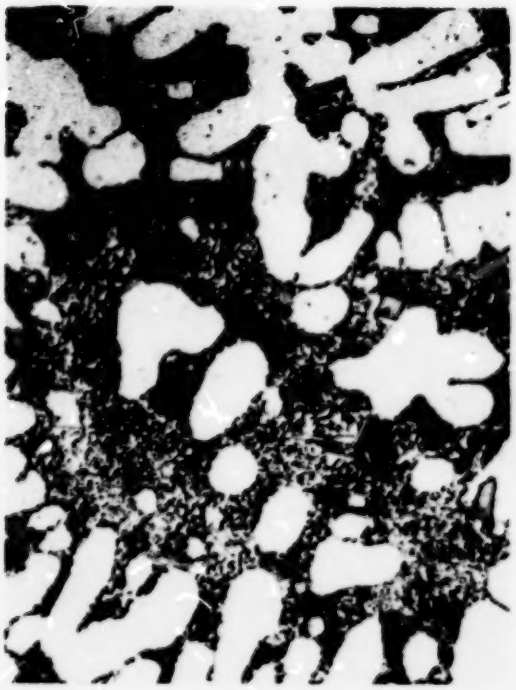


b) 47°C per min
(Plaster Mould with chill)

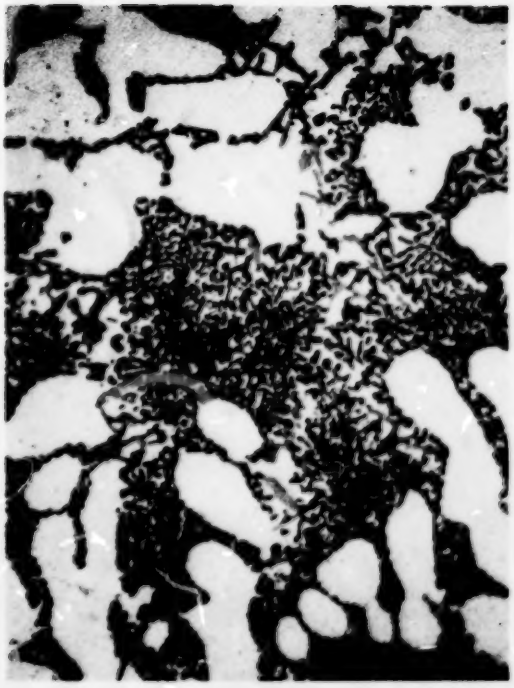
c) 25°C per min (Sand Mould)

d) 14°C per min (Plaster Mould)

FIGURE 9 - EFFECT OF SOLIDIFICATION RATE ON THE MICROSTRUCTURE OF ALLOY A356-T6, X600 mag



a) Steel Mould

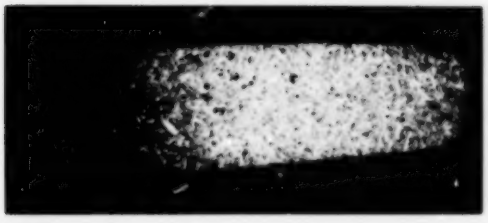


b) Plaster Mould with Chill

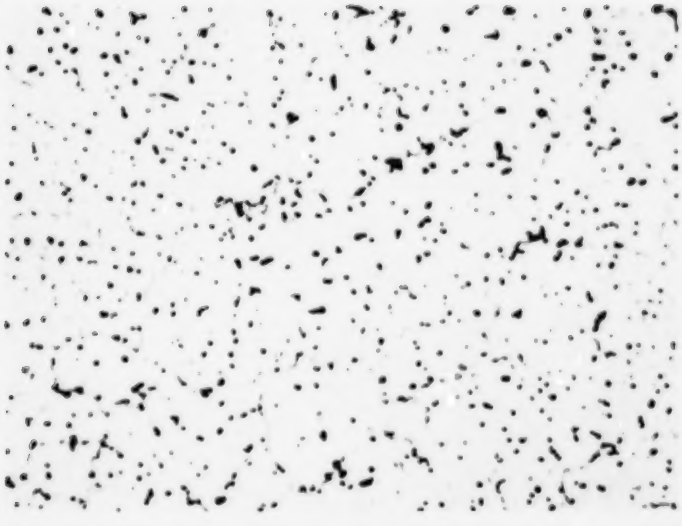
c) Sand Mould

e) Plaster Mould

FIGURE 10 - EFFECT OF SOLIDIFICATION RATE ON THE MICROSTRUCTURE OF ALLOY wa/3.2381 X 600 mag



a) X 3.5 mag



b) X 200 mag

FIGURE 11 - MICROSTRUCTURE OF PERMANENT METAL (STEEL) MOULD CASTING IN ALLOY 535.0

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